# Deformation and Fracture of α-β Brass Two-phase Bicrystals(α-β黄銅二相双結晶の変形と破壊)

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# 論　文　内　容　要　旨

## I　Introduction

To further improve the existing properties of metals and alloys it is mandatory to understand their deformation and fracture characteristics. However, owing to the complexity of studying the deformation and fracture behaviours of polycrystals it is logical to approach the problem through studying bicrystals. Therefore, several researches were previously undertaken on mono-phase bicrystals. However, very little researches were reported about two-phase bicrystals in spite of the current interest. To fill this gap, the present research was undertaken to clarify the deformation and fracture behaviours of the $\alpha$-$\beta$ brass two-phase bicrystals at 150, 300 and 450K.

## II　Experimental procedure

A two-phase bicrystal can be simply defined as a unit consisting of two component-crystals (having different structures) joined together at the interface (see Fig. 1). Such units were prepared in the present investigation by the solid

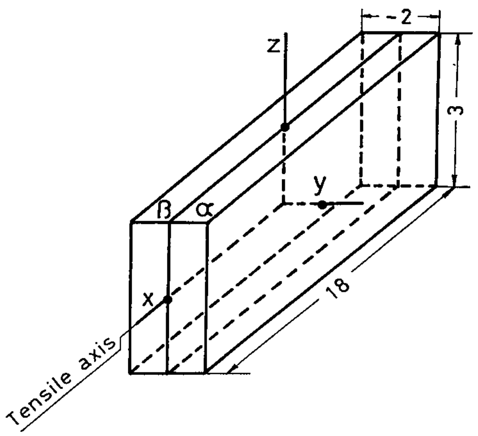


Fig. 1 Bicrystal geometry and gage dimensions(mm).

diffusion technique at 873K. The present bicrystal exhibited the following characteristics ; 1) bulk-like specimens, 2) macroscopic flat phase-boundary, 3) uniform, and thermodynamic equilibrium, concentration profiles in both component phases (as measured by the electron probe micro-analyzer (EPMA)), 4) the orientation-relationship between the two component crystals (as determined by the X-ray back reflection Laue technique) satisfied the plane matching of the close packed planes $\{111\}_\alpha$ and $\{110\}_\beta$. From all the above characteristics it was concluded that the present bicrystals are suitable for the mechanical testing.

## Ⅲ Results and Discussion

To maintain the continuity of material at the interface during the deformation of a bicrystal (see Fig. 1) the following relations should be satisfied ;

$$\varepsilon^{\alpha}_{xx} = \varepsilon^{\beta}_{xx} \qquad \varepsilon^{\alpha}_{zz} = \varepsilon^{\beta}_{zz} \qquad \varepsilon^{\alpha}_{xz} = \varepsilon^{\beta}_{xz} \qquad (1)$$

these three relations plus a specification of the elongation $\varepsilon_{xx}$ will amount four independent conditions for the deformation of bicrystals (four degrees of freedom). Hence, the most incompatible bicrystal will, in general, deform on a total of four slip systems. These four slip systems can be distributed in such a way that there is one slip system in $\alpha$-component and three in $\beta$ or vice versa or two in $\alpha$ and two in $\beta$. It is worth mentioning that the previous analysis holds true for both plastic and elastic incompatibilities.

Owing to the distinctly different elastic properties of both the $\alpha$- and $\beta$-components, which are not the case during the deformation of mono-phase bicrystals, it can be predicted that the elastic incompatibility will play an important part during the deformation of the present bicrystals. Therefore, emphasis has been given to the elastic incompatibility effects. Naturally, the role of the plastic incompatibility plus the effect of the orientation of the component crystals were all taken into consideration. As an example, Fig. 2 shows the slip traces on

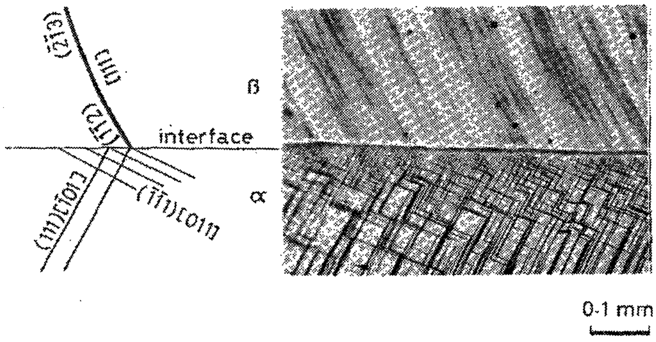


Fig. 2 Slip traces observed on the narrow face of the bicrystal no.1 at 150K (10% plastic strain)

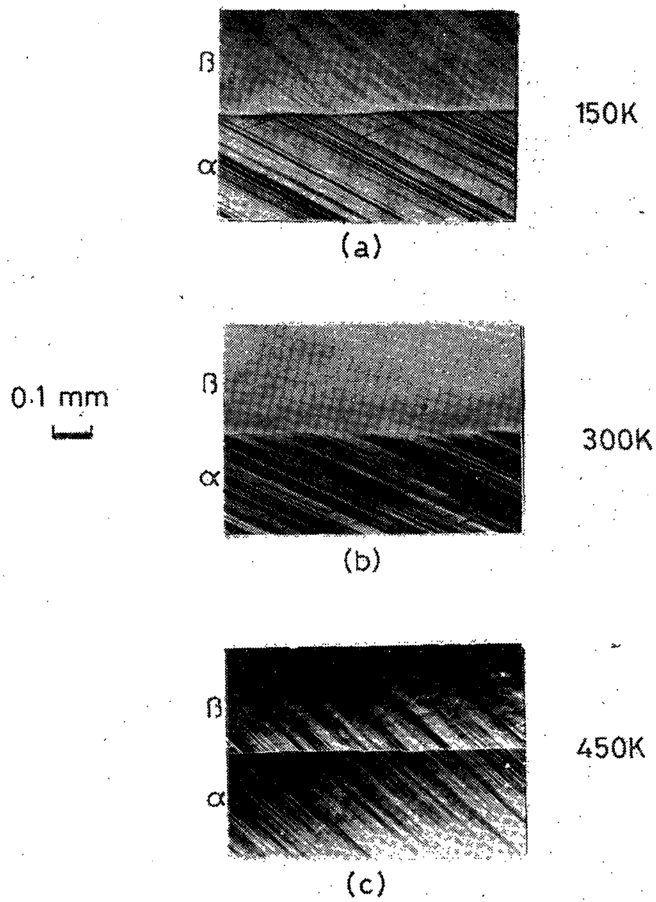


Fig. 3 Slip continuity phenomenon in the $\alpha$-$\beta$ brass two-phase bicrystals.

both phases of a bicrystal tensioned at 150K to a 10% plastic strain. The curvature of the slip traces in $\beta$ is closely related to the elastic incompatibility since it is known to decay exponentially with distance from the interface. The slip systems operating in the interior of $\beta$ is mainly controlled by the Schmid factor and the plastic incompatibility. This curvature in slip traces in $\beta$ was observed at 150 and 300K, while at 450K the slip traces in $\beta$ were always straight due to the change of slip characteristics with temperature. The operative slip systems in $\alpha$ always corresponded to the $\{111\}_{\alpha}\langle 110\rangle_{\alpha}$ type and the slip traces were, at all test temperatures, straight, well-defined and band-like (Cf. Fig. 3). On the other hand, on $\beta$ the slip systems had always the slip direction $[111]_{\beta}$, however, the slip planes lied between $(\bar{2}11)_{\beta}$ and $(\bar{1}\bar{1}2)_{\beta}$. The slip traces characteristics in $\beta$, as can be seen from

Fig. 3, changed with temperature. At 150K, broad band-like, and at 300K fine and ill-defined while at 450K they were straight and rather well-defined and corresponded to the $\{110\}_{\beta}\langle 111\rangle_{\beta}$ type of slip. It is worth mentioning that, slip systems of Schmid factor as low as 0.17 were observed to operate at 450K reflecting the high value of the incompatible stresses.

As can be seen, also, from Fig. 3 while the slip continuity was not observed at all at 150K it was remarkably observed at 450K. A moderate appearance of the phenomenon can be observed at 300K. The phenomenon of slip continuity across the interface is a twofold problem since it depends not only on the slip characteristics of the constituents at each test temperature but it depends also on the misorientation angle between the matching planes (other factors such as the interface structure are asided in the present discussion). In general, the present bicrystal satisfied the plane matching of the $\{111\}_{\alpha}$ and $\{110\}_{\beta}$ which correspond to the slip planes in both phases. However, at 150K slip planes never showed the $\{110\}_{\beta}$ type and therefore never showed slip-continuity. At 450K where the severe dislocation pile-ups are known to occur in $\alpha$, the high stress concentration value which is a prerequisite for the slip-continuity across the interface do exist, additionally, at 450K the slip systems in both phases implies the matching planes. It should also be added that the Burgers vectors of $\alpha$ ($b_{\alpha}$=2.597Å) and $\beta$ ($b_{\beta}$=2.551Å) are known to have a very small difference in magnitude which might result in a very small interface dislocation (deformation ledge) when a dislocation passes from $\alpha$ to $\beta$.

Concerning the stress-strain curves of the present bicrystals, two main parameters were used being ; i) the orientation of the component crystals and consequently the degree of incompatibility of bicrystals, and ii) the test temperature (being 150, 300 and 450K). In respect of the first parameter, it is common to all the curves, at each test temperature, that the bicrystal having a higher degree of incompatibility showed higher values of flow stress and lower values of elongation up to fracture among the other specimens at the same test temperature. Concerning the second parameter, Fig. 4 illustrates a comparison of the stress-strain curves of bicrystals having the same orientation at different test temperatures. The difference among the initial yield stresses, flow stresses and the elongations up to fracture can be plausibly explained on the basis of the different characteristics of the component crystals. Additionally, an interface effect may also be important e. g. at 450K, where slip-continuity was remarkably observed, the flow stress was comparatively low and the ductility was compara-

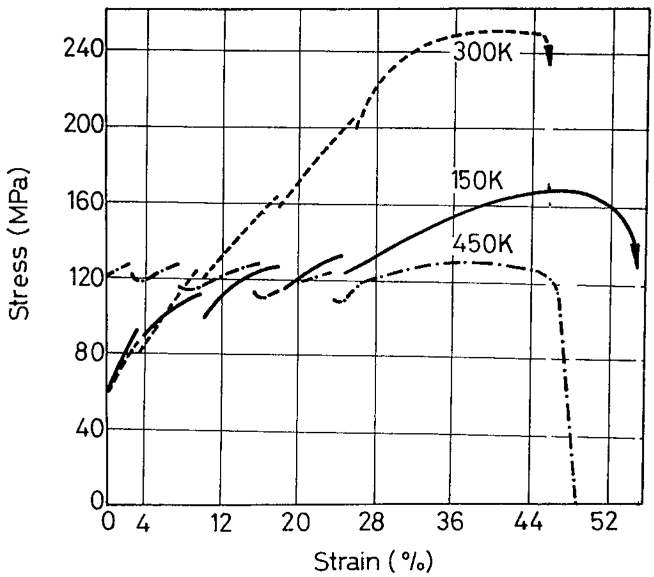


Fig. 4 Stress-strain curves of $\alpha$-$\beta$ brass bicrystals at different temperature.

tively large.

The fracture phenomena were studied in the present bicrystal both statically and dynamically by the use of the scanning electron microscope (S. E. M.). The static observation of the fractured rurfaces revealed, at 150K, shear dimple pattern on both phases. At 300K, $\alpha$ showed a smooth ripple pattern and $\beta$ showed a shear dimple pattern. At 450K the pattern was inverse to that observed at 300K. The change of the fracture pattern due to the change of temperature my be attributed to the change of the characteristics of $\alpha$ and $\beta$ due to temperature and it may also refer to an interface effect. As can be seen from Fig. 5, the dynamic observation at 300K revealed the crack initiation site to be at the interface and the crack propagation direction to be directly related to the orientation of the components.

A supplementary study on the deformation and fracture of $(\alpha+\beta)$ brass polycrystals at 150, 300 and 450K exhibited good agreement with the bicrystal results at the same test temperatures and under similar test conditions.

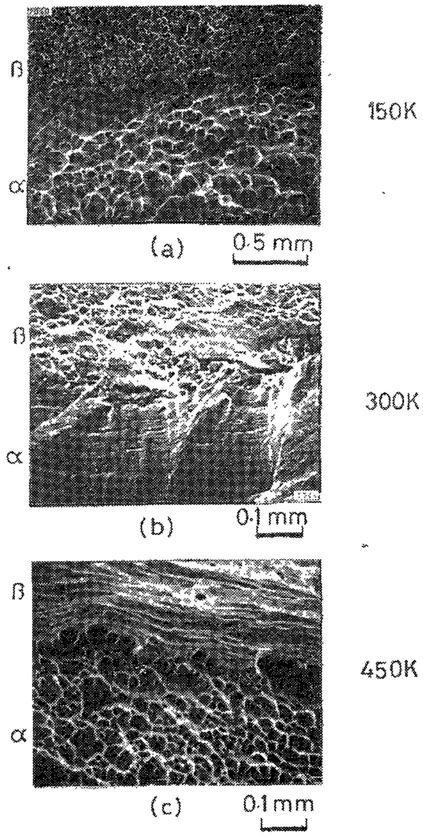


Fig. 5 Fracture patterns of the $\alpha$-$\beta$ brass bicrystals at temperatures.

## Ⅳ Conclusions

$\alpha$-$\beta$ brass two-phase bicrystals obtained by the solid diffusion technique at 873K were found to be suitable for mechanical testing. The deformation characteristics of the present bicrystals were studied at 150, 300 and 450K through studying the slip systems and the stress-strain curves during tension up to fracture. The fracture phenomena were studied by the scanning electron microscope both statically and dynamically. Correlation between bicrystal and polycrystal results could be detected.

# 審 査 結 果 の 要 旨

結晶粒界は金属材料の機械的性質に強い影響を及ぼすため，双結晶を使って粒界挙動を探る研究が盛んに行われているが，その大部分は構成結晶が同一相よりなる単相双結晶に関するもので，異相双結晶に関するものは試料作製が困難なため皆無に近い。しかし現実には複相合金が多く実用されており，異相境界の挙動を明らかにすることは大きな意義を持つ。以上の情勢を踏まえ，著者は新たに開拓した異相双結晶作製法により得た$\alpha-\beta$黄銅双結晶試験片を用い，引張り変形と破壊における異相界面挙動につき系統的検討を試みた。本論文は以上の成果をまとめたもので，全編8章よりなる。

第1章は序論である。

第2章では異相双結晶を固相拡散法により作製する方法の概要を述べ，得られた双結晶の諸特性，すなわち界面形状，溶質濃度分布，および構成相$\alpha$（f.c.c.）と$\beta$（b.c.c.）との方位関係を調べている。本双結晶の構成相それぞれの最稠密結晶格子面である$\{111\}_\alpha$と$\{110\}_\beta$とが，互いに平行関係（Plane Matching）にあることとは興味深い知見である。

第3章は室温（300K）における変形挙動の検討結果である。すべり模様は$\alpha$相では直線状に明瞭に，$\beta$相では曲線状に細かく不明瞭に現われ，これらは界面における弾・塑性ひずみの適合性条件から説明された。

第4章における低温（150K）変形では室温とは異なる挙動が観察されたが，これは構成相自体の低温変形特性によるものであり，また主すべり以外のすべり系の出現は，やはり界面の影響と解釈出来た。

第5章は高温（450K）における結果である。この温度における両相の変形挙動の特異な温度依存性に応じ，すべり線があたかも直線的に界面を横切った観を呈する。しかしすべり系としてはやはり界面効果に起因するものが活動することを明らかにしている。

第6章では破壊挙動を比較検討し，各温度での破面模様の差異を各相自体の温度特性と界面介在による応力状態から説明した。特に走査電子顕微鏡内での破壊の直接観察による界面亀裂発生過程の確認は価値ある知見である。

第7章では双結晶についての結果を多結晶試料についての補足実験結果と対応させている。

第8章は総括である。

以上要するに，本論文は異相境界における変形・破壊現象を，独自の技法で作製した異相双結晶を用いて系統的に解明し，多くの新知見を得ることに成功したもので，金属工学の発展に寄与するところ少なくない。

よって，本論文は工学博士の学位論文として合格と認める。